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Electro-Deposition.

THE art of electro-plating is an old one, dating back to the early days of the last century, but as in so many other industrial arts the progress made in the last twenty years eclipses that of the previous hundred. The discussion which took place at the British Association meeting last year on the initiative of Professor J. C. Philip, President of the Chemistry Section, revealed in a striking way the scientific influence exercised by chemists, physicists, and metallurgists on the development of electro-deposition. As stated in the official report of the 1936 meeting of the British Association, the improvements which have been made in the control of the whole plating process, including the very important preliminary cleaning operations, required to free the surface of the basis metal from traces of grease or other impurities, which would prevent perfect adhesion of the coating, have raised the general standard of quality and reliability of plating to a remarkable extent, and have given users a real confidence in its value for a wide range of new industrial purposes. The resulting increase of demand has stimulated the introduction of mass production methods for plating articles in automatic plants. The articles to be plated are carried at a regular rate by conveyor chains through a succession of tanks and chambers, in which the operations of cleaning, washing, plating, rinsing, and drying are conducted in their proper sequence and each for its appropriate time. Such plants have become a feature of modern electro-plating industry. An automatic nickel-plating plant may carry 3000 to 5000 gallons (or even more) of solution and as much as 5 tons of nickel anodes. It is estimated that the total area covered by nickel plate alone amounts to 250 million square feet per annum. This result has been achieved by the steady expansion of the applications and uses of nickel deposits. To its first application for decorative and protective purposes was added its use in thick layers for building up worn and undersized parts, and later its employment as an undercoat for chromium plating. A high grade of chromium plating on steel cannot be produced without a substantial undercoat of nickel. The latest factor stimulating expansion of nickel plating is the production of bright nickel deposits, still in its early stages in this country, but considerably advanced in America, though the results often leave something to be desired by being, as it has been said, "bright, but not brilliant."

Underlying the whole of this work is a large body of researches bearing on the technical problems of electro-deposition. As Mr. Robert C. Stanley has stated in his review of "The Nickel Industry in 1936," issued by the International Nickel Company: "In Great Britain particularly the standards governing nickel-plated deposits have been improved appreciably, due in large measure to the research work which is being conducted under the auspices of the British Non-Ferrous Metals Research Association. The result is reflected in the 40 per cent. increase in the sale of nickel anodes during 1936 as compared with 1933." It must not be forgotten that for nine years before the British Non-Ferrous Metals Research Association took over the more fundamental work in 1935, the Department of Scientific and Industrial Research, through its Electro-deposition Research Committee, presided over by Dr. W. Rosenhain, undertook the direction of and financial responsibility for an extensive series of researches, carried out at the Royal Aircraft Establishment, Farnborough, the University of Sheffield, and the Research Department, Woolwich, and credit must be given to the Department of Scientific and Industrial Research for this far-sighted policy which has achieved so satisfactory a result.

It is not only in the electro-deposition of nickel and of chromium that advances have been made. In addition to these and to the old-established plating processes for copper, silver, and gold, there have been extensive developments in the electro-deposition of zinc, cadmium, tin, and lead, and of alloy coatings of two or more metals. A non-tarnishable finish on silver has been secured by rhodium plating. In each instance exact correlation of the conditions of deposition with the character and properties of the deposit has been necessary, and the tests for hardness, porosity, internal stress, and cohesion of the deposit, as well as those for the chemical and physical condition of the solutions, have furnished important information which enables the quality of the deposit to be accurately controlled in practice.

All these and many other aspects of the subject are covered by the titles of the papers to be presented at the first International Conference on Electro-deposition, organised by the relatively young but enterprising Electrodepositors' Technical Society, to be held in London on March 3rd and 4th. It will be a truly international conference, for the authors of the papers are drawn from Czechoslovakia (1), France (6), Germany (2), Great Britain (6), Holland (1), the United States of America (10), and the U.S.S.R. (4). The subjects to be dealt with are grouped under the headings: "The Properties of Electro-deposits, Electro-deposition Practice Abroad, Electro-deposition of Precious Metals, and of Base Metals." Many ramifications of the industry, such as electrotyping, electroforming, metal colouring, and anodic oxidation are apparently outside the scope of the programme arranged, but the issues under discussion are of fundamental importance to the industry as a whole, and the activities of the Con-

ference can hardly fail to be of great value in extending the knowledge of the results of research and practice in many countries, and furthering the use of the processes and products of electro-deposition.

Indium.

AMONG the rare metals which are attracting attention on account of their possible practical applications are the related elements of gallium, indium, and thallium. Of these gallium is by far the rarest, and thallium the most common. The possibility of the commercial production of indium is just being realised.

Indium is a soft white metal of low melting point (155 deg. Cent.) and density 7·28. It occurs in numerous zinc blendes and received its name from two strong indigo-blue lines in its spectrum. It is rarely, if ever, present in an ore in amounts exceeding 0·1 per cent., though in the flue-dust from zinc or lead furnaces it may reach 0·2 per cent. Its separation from the other metals present is a complicated chemical process, and the metal is ultimately obtained by reduction of the oxide by hydrogen or sodium, or preferably (since it forms both a hydride and a sodium compound) by electro-deposition from cyanide solutions. Although the metal does not tarnish in air at atmospheric temperature or just above its melting point, it oxidises at red heat and burns with a blue flame. It is slightly oxidised by carbon dioxide above 560 deg. Cent. Its hydride decomposes at 330 deg. Cent. and re-forms on cooling in hydrogen. It is not attacked by alkalis, but dissolves fairly readily in acids. From the time of its discovery in 1863, indium has occupied an important place in theoretical chemistry. For some time it was thought to be a divalent metal (with atomic weight 76·6) on account of its association with zinc. Mendeléef could not fit indium into his periodic table unless it was trivalent (atomic weight 114·8). Later evidence completely supported this view; in its compounds indium displays a valency of one, two, or three. In recent times indium has assumed some importance in the theory of the constitution of alloys. Its structure is face-centred tetragonal, but the axial ratio ($c/a = 1·078$) is so near to 1 that the structure approximates to that of a face-centred cube. With elements of suitable "size factor" (*i.e.*, having atomic diameters within about 15 per cent. of one another), indium should form alloy systems with a certain range of solid solubility. One of the most favourable in this respect is silver. Hume-Rothery has shown, on the basis of his "electron concentration" rule, that the limits of solid solubility for the addition of divalent, trivalent, and tetravalent elements to silver vary as $1 : \frac{2}{3} : \frac{1}{3}$, except in so far as the solubility is also affected by the difference in atomic diameter and lattice distortion of the different systems. When the "size factor" is favourable, the divalent elements have a solid solubility of about 40 atoms per cent.; the trivalent of about 20, and the tetravalent of about 13 atoms per cent. Measurements of interatomic distances of indium compared with that of the adjacent monovalent element silver indicate a state of imperfect ionisation in indium (as also shown by aluminium, thallium, and lead), but in solid solution in silver there is no doubt that indium behaves as a typical trivalent element. Its solid solubility at the peritectic temperature (693 deg. Cent.) is 19·8 atoms per cent., diminishing to 19·4 at room temperature. The corresponding solubility of indium in copper is 11·6 atoms per cent., since the size factor in this case is not favourable.

In addition to the silver-indium and copper-indium diagrams which have been carefully determined, a number of other diagrams are known. Thallium and indium, in which the size factor is favourable and the valencies equal, give an almost complete series of solid solutions with a two-phase field only between 48 and 57 atoms per cent. of thallium. Lithium and sodium give compounds LiIn and NaIn. The lithium compound gives an unbroken range of solid solutions with lithium, but only a limited range with indium. Zinc and indium give a eutectiferous system. The freezing point of the eutectic is 143·5 deg. Cent., and its composition indium 96, zinc 4 per cent. The solid solubility of zinc in indium is practically nil, but that of indium in zinc was not determined. The size factor in this system is not favourable; in the system cadmium-indium it is much more so, also in the systems aluminium-indium and magnesium-indium, none of which has yet been investigated.

Indium is now being recovered in small quantities as a by-product of zinc manufacture in Arizona, and a larger amount is available than ever before. As a result it has passed from the stage of being a laboratory curiosity ten or fifteen years ago to that of a commercial, though still rare, metal. Its price per ounce is said to be about 30 dollars, which is only about one-tenth of its price ten years ago. There is little doubt that if proved fields of utility for indium could be established, its output could be increased.

Considerable exploitation of the metal, based on the investigation of possible uses, is, therefore, going on. Chief among the potential uses are various applications to jewellery and to the production of an untarnishable silver. A study of the silver-indium alloys indicates that 42 per cent. of indium is required in silver to prevent the attack of the alkali sulphides. As alloys with over 40 per cent. of indium are hard and brittle, the high indium content must be applied locally and not throughout the article. Hence, electro-deposition must be used, and since the coating is soft and dull, the object is subsequently heated so that the indium deposit diffuses into the silver. This gives a durable indium-rich surface alloy which will take a high polish; but in this application indium will have a strong competitor in rhodium plating. Special colour effects in jewellery can be produced by indium. Thus, when gold and indium were plated alternately on silver and the article heated, it developed a sky blue colour. Inter-diffusion of layers of silver, indium, and palladium at 816 deg. Cent. yielded a fine rose-pink alloy. Other applications of indium are dependent on the low melting point of the metal. The fusible alloy of lowest melting point generally available is Lipowitz's alloy (bismuth 50, lead 27, tin 13, cadmium 30), having a melting point of 70 deg. Cent. (158 deg. Fah.). No method of appreciably reducing this temperature was known apart from the use of mercury. It is found, however, that the addition of indium to Lipowitz's alloy in the proportion of 18 per cent. of indium to 82 per cent. of the alloy, reduces the freezing point to 46·5 deg. Cent. (116 deg. Fah.), which is not far above the normal temperature of the body. In general properties, the alloy resembles Lipowitz's alloy, but owing to the lower temperature involved it oxidises less in the fluid condition. The fact that the human

body can come in contact with the liquid alloy without discomfort suggests the possibility of its use for casts and impressions. Such casts can be made permanent by electro-deposition of copper or silver on them and subsequent melting out of the fusible alloy. Cloth or blanket impregnated with the alloy may be moulded to any form when heated above 116 deg. Fah., and when cooled sets to a rigid condition. When it is desired to remove the cast, hot water bottles may be applied so that the material again becomes pliable. The alloy may thus serve as a substitute for plaster of paris in surgical work. It may also have applications in dental work.

The commercial future of indium is dependent on the discovery of some applications which will create a demand for the metal, but much investigation and development work will be required to achieve that position. It appears that a moderate demand, if it arose, could readily be met.

The Alloys of Iron and Zinc.

By W. D. JONES, M.Eng., Ph.D.

THE alloys of iron and zinc have always been of technical interest in view of their relation to the structure of galvanised coatings, attack of galvanising pots, and sheradising. Recently, however, these alloys would appear to be becoming an even greater centre of interest from a theoretical standpoint.

An account of the iron-zinc equilibrium diagram has been given in this journal by Jones and Parker.[1] In the present article only data which have subsequently been published will be considered.

The whole of the iron-zinc system has now been reconsidered by Schramm[2] and some investigational technique has been devised which is worthy of special attention. The systems of the iron-zinc, iron-antimony, iron-arsenic types present two-fold difficulties: (i) the intermediate intermetallic phases are fragile and invariably lead, on cooling and sectioning the alloy, to a porous and sometimes powdery mass which frequently contains much oxide; (ii) on account of the volatility, in each case, of the non-ferrous metal, the high iron alloys of desired composition are difficult to prepare.

The first problem in the case of the iron-zinc alloys is encountered in those alloys containing between 80 and 96 per cent. of zinc. Schramm has devised an elegant means of avoiding this trouble. The melts, on cooling to 900 deg. Cent., are introduced into a crucible containing molten salt, which freezes before the separation of the intermetallic compound. In this manner, entrance of air is prevented whilst the alloy is unable to expand. The only possible alternative to this simple technique would involve the casting of the melts in a somewhat elaborate apparatus employing an inert atmosphere and pressure.

In dealing with the experimental difficulties of the iron end of the system, Schramm has employed the powder technique evolved by Jones[3] in similar investigations. Schramm prepares the alloys containing less than 80 per cent. of zinc by mixing together a powdered 85/15 iron-zinc alloy with ferrum reductum, or carbonyl iron, pressing at approximately 20 tons per square inch, heating at 300 to 400 deg. Cent. in hydrogen, followed by longer annealing at higher temperatures in air-tight glass containers. In view of the frequent impurity of both of these types of iron, together with the unimportance of fine powders in such alloys, Jones employed filings of zinc and electrolytic iron, pressed at 35 tons per square inch in a chromium-lined steel bomb coated internally with cement, which, when plugged up, was used as the annealing container. Both methods are apparently satisfactory and are examples of the utility of the powder technique in approaching such alloys.

The diagram evolved by Schramm is illustrated in Fig. 1. It differs from the findings of Jones in respect of its retention from the older diagrams of the eutectoid at approximately 27 per cent. of zinc. If zinc

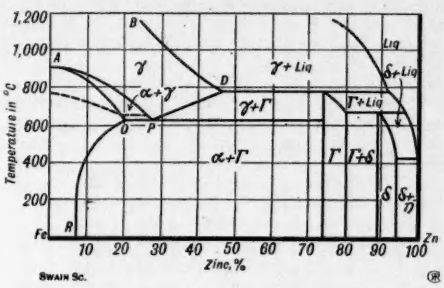

FIG. 1—IRON-ZINC DIAGRAM (SCHRAMM)

is allowed to diffuse into iron at temperatures between 900 and 1400 deg. Cent. a diffusion boundary is obtained, and this has been shown by Jones to be indicative of a phase change and has been associated with a gamma loop. The new diagram provides no explanation of this phenomenon. Whilst sufficient evidence appears to have been provided for the lines R O, P D, and D B, and whilst an apparently genuine eutectoid illustrates the work, it must be admitted that the evidence for the lines A O and A P is still insufficient.

The number of the intermetallic phases in the iron-zinc system seems to be decreasing. In the previous review it was concluded that positive data existed for two compounds, whilst there was considerable evidence for a third. Such evidence is obtained by a consideration of the structures obtained by exposing iron to molten zinc at various temperatures. In such structures three, if not four, layers may be distinguished, although one may be a mixture of two phases. Rigg[4] seems to have demonstrated the existence of three compounds by diffusion between iron and solid zinc. Until such structures have been explained by the diagram, and both diagram and structures correlated in the manner employed by Jones and Hoare[5] for iron-tin alloys, it cannot be said that the equilibrium diagram is in a very sound position.

Schramm finds two compounds which are described as follows:—

(i) δ between 88·5 and 93·7 per cent. zinc, hexagonal, usually known as $FeZn_7$;

(ii) Γ between 73 and 80 per cent. zinc, cubic with 52 atoms in the unit cell, previously known as $FeZn_3$, Fe_5Zn_{21}, or Fe_3Zn_{10}, but not corresponding exactly with any of these.

It will be noted that the conclusions on this latter compound are quite at variance with those of Stillwell and Clark[6] and Osawa and Ogawa,[7] which were reported in our previous review. Both of these investigations reported Γ to be body centred. Not the least disturbing feature of the iron-zinc alloys is

the fact that each fresh investigation yields data, not in the nature of an adjustment or confirmation, but rather pointing to totally new viewpoints. The remaining features of Schramm's diagram are apparent in Fig. 1.

Perhaps a clue to these conflicting investigations might be found in the recently published work of Scheil[8] and of Püngel, Scheil, and Stenkhoff.[9] The rate of attack of zinc on various steels was investigated at various temperatures by studying the loss in weight of the steel after simple immersion in molten zinc. The results for various steels are given in a table and are illustrated here, not only because of their theoretical interest, but partly because of their technical utility. It will be noticed that the steels have been divided sharply into two groups (steel 3 being in both), I where the rate of attack is very low, and II very high. A detailed analysis of these results will not be given here, but it is immediately noticeable that some relation appears to exist between the copper and carbon contents with group II and chromium and aluminium for group I. No explanation of this behaviour has been proffered.

Analysis of Steels and Rate of Attack Immersed in Zinc.

Steel No.	Composition, per cent.												Group.	Max. rate of attack.	
	C.	Si.	Mn.	P.	S.	Ni.	Cr.	Cu.	Al.	Mo.	Ti.	Sn.		gm² h.	At °C.
1	0·03	—	—	0·003	0·004	—	—	—	—	—	—	—	I	low	—
2	0·10	0·02	0·65	0·030	0·048	—	Trace	0·19	—	—	—	—	II	900	480
3	0·42	0·18	0·69	0·021	0·044	—	—	0·11	—	—	—	—	I/II	280	480
4	1·00	0·26	0·51	0·024	0·012	—	Trace	0·15	—	—	—	—	II	1320	460
5	0·26	0·01	0·65	0·102	0·042	—	—	1·04	—	—	—	—	II	1000	480
6	0·16	0·25	0·90	0·024	0·028	—	0·43	0·72	—	—	—	—	I	low	—
7	0·04	0·05	0·20	0·019	0·032	—	1·47	0·14	—	—	—	—	II	840	440
8	0·07	0·12	0·18	0·016	0·027	—	4·48	0·11	—	—	—	—	I	low	—
9	1·07	0·17	0·26	0·010	0·028	—	1·47	—	—	—	—	—	I	low	—
10	0·04	0·06	0·44	0·026	0·036	—	—	0·16	0·05	—	—	—	II	640	460
11	0·04	0·08	0·18	0·026	0·029	—	—	0·27	0·27	—	—	—	II	580	440
12	0·03	0·11	0·28	0·022	0·029	—	—	0·27	0·47	—	—	—	II	830	440
13	0·05	0·14	0·34	0·022	0·036	—	—	0·12	1·76	—	—	—	I	low	—
14	0·10	0·45	0·30	0·020	0·002	—	12·0	—	3·00	0·50	—	—	I	low	—
15	0·04	0·17	0·23	0·024	0·029	—	—	0·41	0·50	—	—	0·38	II	480	440
16	0·03	0·24	0·30	0·031	0·029	—	—	0·24	0·55	—	—	0·14	II	660	440
17	0·04	0·12	0·16	0·035	0·026	0·95	—	0·40	—	0·44	0·21	—	I	low	—
18	0·04	0·10	0·26	0·031	0·022	0·95	—	0·33	—	2·86	0·29	—	I	low	—
19	0·06	0·24	0·49	0·003	0·023	—	—	0·04	—	1·59	—	—	I	low	—

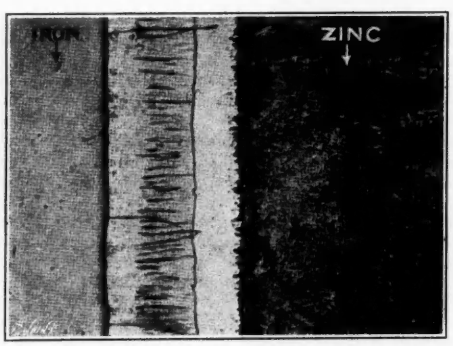

FIG. 2

A microscopic examination of the reaction products of these steels is most interesting. Group I shows three compound layers typical of electrolytic iron (Fig. 2). These layers for the purposes of discussion only will be designated $FeZn_3$, nearest steel; $FeZn_7$, nearest zinc; and in between a mixture of both compounds. In the group II steels (Fig. 3) only one phase ($FeZn_7$) is apparent, and the greatly increased rate of attack is indicated by the much higher concentration of compound in the zinc. In some specimens (steel 3, for example) both types of attack were perceivable. No explanation of these phenomena is forthcoming, but their importance in connection with the interpretation of reaction structures and the diagram, and with the use of such steels for galvanising and galvanising pots is immediately apparent.

By continuously raising the steel specimen (group II) out of the zinc bath whilst the compound is growing, layers of "$FeZn_7$" of relatively enormous size can be produced. Fig. 4 illustrates such an example and gives an indication not only of the rapid rate of growth under these circumstances, but also of the force of crystallisation of the compound.

Musatti and La Falce[10] have also studied the rate of attack by zinc of a selection of steels. This work has been approached with the technical utility of the results in mind. Whilst no such phenomenal rates of growth are reported in this investigation, it is interesting to note that photomicrographs of both

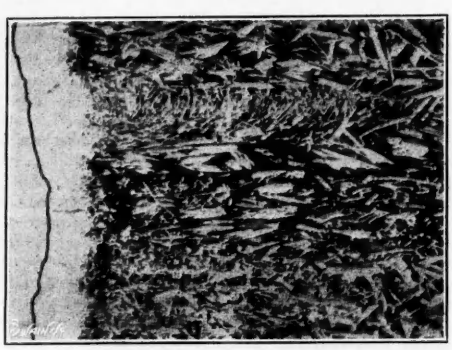

FIG. 3

types of structures groups I and II are reproduced, II being high-carbon steels, and I medium-carbon. These investigators also report the influence of manganese, silicon, nickel, chromium, sulphur, and phosphorus on the rate of attack, but most unfortunately do not mention copper or aluminium. The paper includes figures of practical utility and should

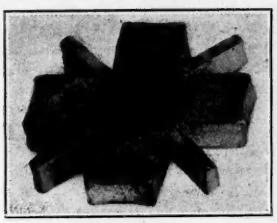

FIG. 4

be consulted if information in this direction is required.

These various investigations into the abnormal behaviour of certain steels provide rather startling evidence in support of the often repeated warnings in connection with the possibly vitiated results obtained by using impure materials in the investigation of a system.

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Carbonyl Steels.

THE production of iron by the formation and decomposition of iron carbonyl on a small scale has been succeeded by attempts to convert carbonyl iron into castings or ingots of iron and steel by re-melting or sintering. It is clear that sintering, if equally effective, would be a cheaper process than re-melting. A paper by E. K. Offermann, H. Buchholtz, and E. H. Schultz* on "The Production and Properties of Steel Made from Carbonyl Iron" describes attempts to produce large ingots of carbonyl iron and carbonyl steels by sintering.

The carbonyl iron powders available contained :—

Carbon, per cent.	Oxygen, per cent.
1 to 1·4	1·0 to 1·4
1·6	4·3
1·9	1·1

From a suitable mixture of the first and second, sintered at 1200 deg. Cent. for six hours, ingots 625 mm. square in section were made and rolled into billets 80 mm. by 80 mm. in section. The carbon distribution was, however, very far from being uniform. Instead of being a medium carbon steel throughout, the ingot actually contained 1 to 1·2 per cent. of carbon at the core and 0·15 to 0·27 per cent. at the outside. Similarly ingots intended to be very low in carbon were found to contain 0·02 per cent. at the outside, but as much as 0·36 per cent. at the core. Tests were therefore made with small ingots, 15 kilos. in weight, and these were more successful. One of them, 150 mm. in diameter, sintered for three hours at 1050 deg. Cent. and subsequently forged to 30 mm. square, was intended to be pure iron; it contained 0·015 per cent. of carbon, 0·017 per cent. oxygen at the skin, and 0·02 to 0·03 per cent. carbon, 0·023 per cent. oxygen at the core. The nitrogen content of the outer region, originally 0·003 per cent., increased considerably with time and temperature of the sintering process. Steels of higher carbon content were more difficult to obtain in a uniform condition. The heterogeneity was not avoided by long sintering time or by subsequent annealing. It was due to lack of uniformity in the progress of the chemical reactions leading to the elimination of carbon and oxygen from the carbonyl iron powders. Further investigation showed that it could be avoided by :—

(1) Reducing the iron powder to the finest possible state of division and thoroughly incorporating it with carbon black by longer grinding in ball mills.

(2) The addition of phosphorus and sulphur in quantities usually found in steel.

(3) The use of graphite as an addition.

The production of larger ingots was then again attempted from the first powder with graphite and ferro-manganese, in quantities adjusted to give a steel containing 1 per cent. of carbon and 0·3 per cent. of manganese. This was sintered for six hours at 950 deg. to 1000 deg. Cent. in a round mould 250 mm. in diameter and forged to 50 mm. square. The chemical composition was almost uniform over the cross section. Steels with 0·1 and 0·35 per cent. carbon (without manganese) and an iron of high purity were also made. The three steels, carbon 0·1, 0·35, and 1 per cent., were then compared with commercial steels of approximately equal carbon content.

In structure the carbonyl steels had the coarsest grain size and the pearlite showed the greatest tendency to granular form. This was checked by the presence of manganese. Hardly any non-metallic inclusions were found in the carbonyl steels which had been sintered without the addition of manganese. Inclusions were considerable in the sintered steel made with ferro-manganese, but in similar re-melted steel they were small in amount. Oxygen content was not higher in carbonyl steels than in the others, and the rate of solution in hydrochloric acid was very low for specimens free from inclusions, *i.e.*, those made without admixture of ferro-manganese. Electrical resistivity was lower in carbonyl steels than in other steels, even after corrections had been applied for impurities present, probably on account of the form of the pearlite. The composition of the carbonyl steels and their tensile properties (read from graphs in the original paper) are given in Table I. The sintered steels showed a lower yield point and ultimate stress than commercially melted steels with approximately equal elongation and reduction of area. This may again be explained by the granular character of the pearlite. The impact value of carbonyl iron (0·02 per cent. carbon) in the annealed condition is

* *Stahl und Eisen*, Sept. 17th, 1936, 56, 1132 (with English translation).

very low (0·5 mkg./cm.²) owing to the presence of carbide in the grain boundaries. After quenching and tempering it was good (24 mkg./cm.²).

TABLE I.—*Composition and Tensile Properties of Sintered Carbonyl Iron and Steels.**

	C.	Si.	Mn.	P.	S.	N.
1. Carbonyl iron	0·02	0·01	0·00	Tr.	0·004	0·022
2. Carbonyl steel	0·11	0·01	0·00	0·011	0·002	0·020
3. Carbonyl steel	0·34	0·00	0·00	Tr.	0·002	0·020
4. Carbonyl steel	0·98	0·00	0·33	0·005	0·004	0·018

	Yield point.	Ultimate stress.	Elongation per cent. (L = 5 *d*).	Reduction of area, per cent.	Impact value, mkg. per cm.².
	Tons per sq. in.				
1.	12	19	40	81	24
2.	14	24	38	75	21
3.	16	30·5	30	56	10
4.	22	42	19	28	2

* Condition not specifically stated ; probably normalised.

The good welding properties of the softer carbonyl steels was confirmed. Case-hardening tests showed a lower depth of carburisation than in ordinary steels. Hardenability and depth of quench-hardening lay at the lower limit of the range of scatter of commercial steels. In this respect, as also in its tensile properties, re-melted carbonyl steel was rather better than the sintered variety (analysis showed that 0·08 to 0·18 per cent. of silicon was taken up on re-melting). Resistance to wear of carbonyl steels, as measured by a laboratory test, was similar to that of other steels of equal tensile strength.

The Austenite-Troostite Change in Relation to Mass-Effect.

THE choice of a suitable composition of alloy steel for a particular purpose involving treatment of a mass of a certain size is largely determined by the "mass effect" in the heat treatment of the steel. The influence of mass is generally determined by a series of direct experiments on samples of different sizes, but the need for a more fundamental method of approaching the question has long been felt. This has given rise to the importance attached to critical rates of cooling, *i.e.*, rates of cooling just sufficient to cause full hardening. When the rate of cooling of a steel is quickened, the carbon change point Ar_1 which is associated with the austenite→pearlite change is depressed and gives two arrests Ar′ and Ar″. On quenching still faster the Ar′ point is entirely suppressed and only Ar″ remains. At Ar′ austenite transforms to troostite, at Ar″ to martensite. Hence, the critical rate of cooling might be defined as that at which Ar′ and troostite no longer appear, but at which the γ–α change takes place at Ar″ and the structure is wholly martensitic.

For practical purposes the use of the critical rate of cooling as a criterion for depth-hardening is open to some objections. Many steels show a gradual increase of hardness to a maximum as the rate of cooling becomes faster. The stages of the transformation of austenite overlap. Troostite and martensite may occur in varying proportions, and the further stage (retention of austenite) may in some degree exist in a specimen most of the austenite of which has transformed at Ar′. In practice, moreover, the essential requirement is not that a martensitic structure should be produced throughout the mass by cooling through the critical range, but that a uniform fine structure, from which massive ferrite and massive carbide are both entirely absent, should be obtained for subjecting to the further tempering treatment.

Uniform hardening thus requires that the austenite transformation throughout the whole cross section should be delayed to the Ar′ point, *i.e.*, until the austenite transforms to troostite, and that in order to correct for local differences in the extent of the Ar′ change and temperature at which it occurs, the subsequent tempering must be carried out above this temperature. Such circumstances would ensure that the mechanical properties at all parts of the cross section are the same or approximately the same.

This condition was put forward as essential for uniform hardening by H. Krainer,* who has studied the transformation occurring at Ar′ in a chromium-molybdenum steel. The austenite → martensite change at Ar″ was not examined. E. C. Bain† has shown that the Ar′ change, when once started, proceeds as a typical first order reaction in which the rate of change at any moment is proportional to the amount of the unchanged austenite, but that the transformation is delayed by a period of incubation,

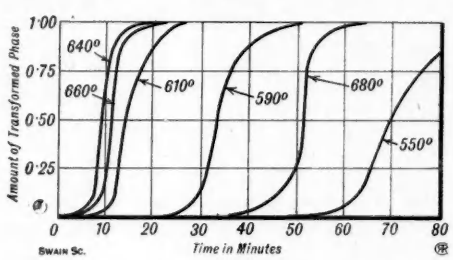


FIG. 1—*Progress of Austenite Transformation at Different Temperatures. Chromium-Molybdenum Steel, Initial Temperature, 850 deg. Cent.*

or nucleation and only attains its maximum velocity after 20 or 30 per cent. of the austenite has been transformed. This was demonstrated with a steel containing carbon 0·55, and manganese 3·8 per cent. at 315 deg. Cent.

Krainer, working with a steel containing carbon 0·5, chromium 1·1, and molybdenum 0·25 per cent., used rods 10 mm. diameter by 60 mm. long, heated in an electric tube furnace to the required initial temperature and held there for a given time, then quenched in a metal bath at the transformation temperature to be studied and quickly transferred to a furnace held at the same temperature and equipped for magnetic measurements. In the first series of experiments the specimens were heated at 880 deg. Cent., and the course of the transformation followed at constant temperatures between 550 deg. and 680 deg. Cent. Typical curves are shown in Fig. 1. They take the ideal form (illustrated in Fig. 2) of a first order reaction preceded by a period of induction or nucleation (*Keimzeit*). The period of induction is of much greater significance in the hardening operation than the velocity coefficient of the reaction.

* *Archiv für das Eisenhüttenwesen*, June, 1936, 1935/36, 9, 619.

† "On the Rates of Reactions in Solid Steel," "Trans.," A.I.M.E., 1932, 100, 13.

Initial temperature is known to have a marked effect on the critical cooling velocity of some steels. This was investigated for changes occurring at 640 deg. Cent. in specimens which had first been heated for 15 min. at different temperatures between 850 deg. and 1100 deg. Cent. (Fig. 3) or held for various times at 850 deg. Cent. (Fig. 4). Both higher temperature and longer time give rise in the steel used to a longer induction period. The temperature to which the austenite transformation must be lowered in order to obtain uniformity throughout the section was

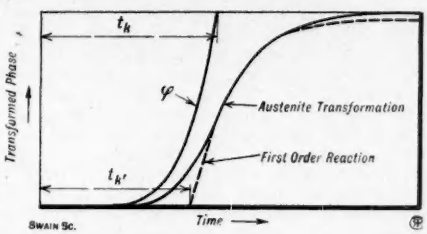


FIG. 2—*Relation between Curves Representing a Reaction of the First Order, and the Austenite Transformation at* Ar′. *(φ indicates progress of formation of nuclei, and t_K′ the period of nucleation.)*

found to be lower the higher the desired mechanical properties.

The properties are judged by tensile strength only. In direct measurements of mass effect, other properties are taken into account, and if this had been done it would probably have been found that the bare attainment of the necessary tensile strength is not sufficient to ensure the best combination of properties. If instead of employing the minimum possible lowering of Ar′, the temperature of the transformation is still further depressed, and the tempering temperature raised, the same tensile strength may be obtained with a better limit of proportionality and yield ratio, and

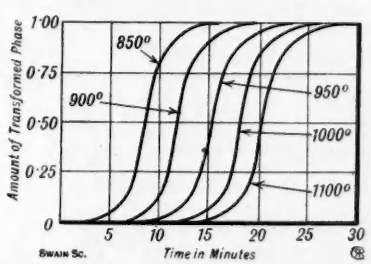


FIG. 3—*Influence of Initial Temperature on the Progress of the Austenite Transformation at* 640 *deg. Cent.*

probably a better reduction of area and impact figure also.

To make use of such studies of the austenite-troostite transformation in connection with mass effect it is necessary to be able to forecast the behaviour of steel cooling at a given velocity by the aid of a knowledge of its behaviour at a succession of constant temperatures. At a constant temperature the transformation will not occur until the induction period at that temperature has been exceeded; in other words, the quotient, time held at a given temperature divided by induction period, must always be less than 1. During cooling, according to Krainer, the austenite-pearlite transformation will be suppressed if the sum of these quotients for all temperatures is less than 1. Unfortunately difficulties arise in any attempt to integrate these quotients because the period of induction is not a simple function of the temperature. Hence it seems unlikely that any numerical factor of relative hardenability is going to be obtained by applying this method. Nevertheless,

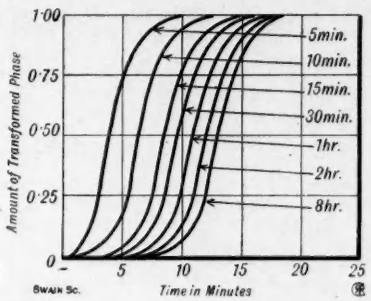


FIG. 4—*Influence of Duration of Heating at* 850 *deg. Cent. on the Progress of the Austenite Transformation at* 640 *deg. Cent.*

transformation curves of various steels involving determinations of the periods of induction are likely to give a basis for a less empirical study of mass action than has usually been attempted.

Phosphorus in Steel.

A JOINT report of the Steelworks Committee and of the Committee of Materials of the Verein deutscher Eisenhüttenleute was published a few months ago in the form of a paper by A. Ristow, K. Daeves, and E. H. Schulz[1] on "The Effect of Phosphorus on the Properties of Basic Plain Carbon Steel." This paper adds one more to the formidable list of recent researches dealing with the effect of phosphorus on steel. The prolonged investigation of a Committee of the American Society for Testing Materials on the effect of phosphorus and sulphur in steel and wrought iron led to the conclusion that in relatively soft homogeneous steels, phosphorus up to 0·1 per cent. has a decided strengthening and only a mild embrittling effect, but that phosphorus in higher concentration, such as might be locally obtained in segregated low phosphorus steels, may cause embrittlement. This Committee has just published its final report (which deals with the effect of sulphur), and since this completes the current programme of work the Committee has been disbanded.

Several recently introduced low-alloy high-strength steels contain about 0·1 per cent. of phosphorus. Professor Andrew and Mr. D. Swarup,[2] as a result of their work carried out for the Alloy Steels Research Committee, came to the conclusion that, provided the carbon content is not greater than 0·15 per cent., steels containing as much as 0·15 per cent. of phosphorus can, by the addition of small amounts of alloying elements, be so treated as to give material possessing not only excellent tensile properties, but also a high impact value and improved resistance to

[1] *Stahl und Eisen*, August 6th and 13th, 1936, 56, 889.
[2] First Report of the Alloy Steels Research Committee, Iron and Steel Institute, 1936, page 227.

corrosion. In particular, the elements chromium and copper, which in the amounts used had no considerable effect on the tensile properties, greatly enhanced the impact value. The work of Andrew and Swarup further suggests that considerable improvement in properties results from the addition of 0·2 per cent. of aluminium, possibly on account of its effect in refining the grain size and increasing the dispersion of the constituents of the sorbitic pearlite formed. The investigators hold that steels containing a reasonably high content of phosphorus may be used profitably where higher tensile properties than those of the low phosphorus steels are required; in other words, that phosphorus must be regarded as a useful strengthening and alloying element and not merely as an impurity.

The German report referred to at the beginning

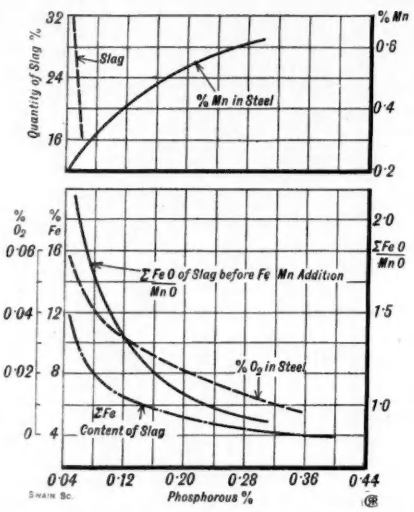


FIG. 1—*Relation between the Phosphorus Content of Basic Bessemer Steel with Less than 0·1 per Cent. Carbon and the Manganese and Oxygen Content of the Steel, and also the Quantity and Composition of the Slag.*

of this article takes account of the economic advantages of high phosphorus content in basic steel as well as its effect on the properties of the finished steel. The first part of the paper deals mainly with steelworks practice, and gives evidence to support the conclusion that in the basic open-hearth and the basic Bessemer process, a reduction of phosphorus in the finished steel means an increase of oxygen in the slag and in the steel. The most efficient and most frequently applied method for the removal of phosphorus is the use of a higher iron oxide content in the slag. This results in an oxygen content in the steel which must be considered injurious. The next most important means of dephosphorising is the use of a higher lime content or increased basicity of the slag. This leads to a great increase in the amount of slag and to an excessive and useless rise in its iron content, probably through the formation of calcium ferrite. On account of the oxygen present economy in the use of ferro-manganese is very unfavourably influenced by a low phosphorus content in the finished steel, and the proportion of added manganese passing into the steel is much reduced. These effects, as they apply to basic Bessemer steel, are illustrated in Fig. 1.

In dealing with the influence of phosphorus on the mechanical properties the authors of the report find that correlation of the published work of a number of investigators confirms the well-known effect of phosphorus in raising the tensile strength and elastic limit and in reducing elongation. In low carbon steel (carbon under 0·1 per cent.) this effect is relatively more marked for the first 0·1 per cent. of phosphorus than for successive additions of 0·1 per cent. The effect of increasing phosphorus on the impact figure is best studied by a series of tests made at different temperatures over a short range above and below atmospheric. Such a series of tests on low carbon steels is represented in Fig. 2. It will be recalled that G. R. Bolsover[3] used this method in investigating various aspects of brittleness, including the effect of phosphorus on strain-age-embrittlement. The detrimental effect of phosphorus in this connection is not mentioned in the German report[4]. On the other hand, in several respects advantages were claimed for a relatively high phosphorus content. For example, a low phosphorus content was held to give rise, directly or indirectly, to an increase in non-metallic inclusions (consequent on the increased oxygen content of the steel) and to such defects as bad weldability, poor wearing properties (for example, in rails), and the sticking of thin sheets on rolling. The influence of

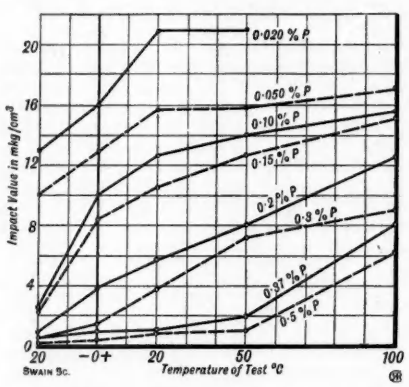


FIG. 2.—*Notched-bar Impact Figure of Killed Basic Open-hearth Steel with Less than 0·14% per Cent. Carbon, in the Rolled Condition, at Different Temperatures. (Test Pieces, 10 mm. × 10 mm. × 60 mm., notch 2 mm. diameter, 3 mm. Deep.)*

phosphorus content on the occurrence of "stickers" is shown in Fig. 3, which represents data from 4300 sheets, 0·22 mm. thick. The best results were obtained with steels containing 0·08 per cent. of phosphorus. An increase in phosphorus from 0·04 to 0·045 per cent. is claimed to have a marked effect. Taken in conjunction with the fact that the lower the phosphorus the higher the oxygen content, the results are held to confirm those of W. Titze,[5] who found that an oxygen content of over 0·015 per cent. was apt to cause sticking, and that the higher the oxygen the worse the sticking became.

The deep drawing properties as measured in the

[3] "Brittleness in Mild Steel," *Journ.*, Iron and Steel Inst., 1929 (1), 473.

[4] See also A. Kuhle, *Metals and Alloys*, 1929, 1, 172. Andrew and Swarup found that phosphorus did not increase susceptibility to quench-age-embrittlement, but strain-age-embrittlement is not dealt with in their report.

[5] *Stahl und Eisen*, 1929, 49, 897.

Erichsen machine on sheet 0·28 mm. thick were stated not to be impaired, but slightly improved, by a rise of phosphorus from 0·04 to 0·15 per cent.

Machinability was improved by the presence of 0·1 per cent. phosphorus though the real free cutting steel requires 0·2 to 0·25 per cent. of sulphur as well. There was stated to be considerable evidence that weldability was increased, and a rise of phosphorus content had a marked effect in reducing wear as measured by a Spindel or an Amsler machine.

The atmospheric corrosion of high phosphorus steels is briefly discussed on the basis of previous experiments by one of the authors, K. Daeves, who had observed in steel with 0·2 to 0·3 per cent. copper and a phosphorus content of 0·06 per cent. appreciably less atmospheric corrosion than in steel with

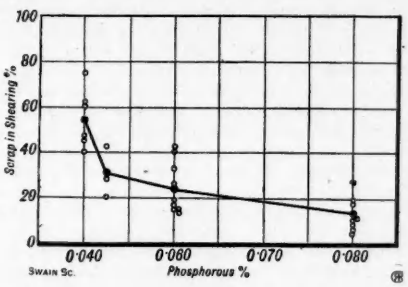


FIG. 3—*Influence of Phosphorus Content on the Amount of Material Scrapped because of Sticking in thin Sheets. Each Point Represents the Mean of* 160 *Sheets.*

the same copper content and low phosphorus. Daeves, Schultz, and Carius have shown that the protective layer on copper steels of equal copper content is thicker in the presence of high than of low phosphorus. It is accepted by the authors that in a plain carbon steel, an increase of phosphorus from 0·01 to 0·1 per cent. has no effect on atmospheric corrosion, but that in steel with 0·15 per cent. of copper the resistance to corrosion was considerably increased by raising the phosphorus 0·1 per cent.

The desirability of alloy additions (fully borne out by Andrew and Swarup's extensive atmospheric corrosion tests) is supported in the German report by reference to the United States Steel Corporation's product Cor-ten, stated to contain copper 0·3 to 0·5 per cent., phosphorus 0·1 to 0·2 per cent. This would imply that Cor-ten derived its properties from the copper and phosphorus alone, whereas the silicon and chromium which it contains must have a very marked influence.[6] An interesting feature of the discussion was the claim made by R. Hilger that a marked increase in fluidity of the steel accompanied a rise in phosphorus from 0·05 to 0·15 per cent. The compositions required were obtained by adding ferro-phosphorus to open-hearth and electric furnace steels, and fluidity estimated from the length of rod (10 mm., 7 mm., 5 mm., and 3 mm. diameter) run from a common runner. In the 5 mm. and 7 mm. diameter rods the length run was almost doubled by this increase in phosphorus, but with more than about 0·15 per cent. showed a tendency to fall off again.

[6] The composition of Cor-ten is C, 0·10; Mn, 0·1 to 0·3; P, 0·1 to 0·2; S, 0·05 max.; Si, 0·5 to 1·0; Cu, 0·3 to 0·5; Cr, 0·5 to 1·5 (J. C. Whetzel, "Modern Steels," American Iron and Steel Institute, May, 1935).

This report serves to focus attention on the wide range of evidence which goes to show that the properties of low-carbon steels are not, for most purposes, unfavourably influenced by the presence of a high phosphorus content, but that this element may, in conjunction with other alloy additions, confer desirable properties. There is as yet, however, no evidence that it would be safe to raise the permissible phosphorus content of medium or high carbon steels, such as rail or spring steel or steels heat treated for use in a hard condition. It is, in fact, only in steels of very low carbon content that phosphorus may be allowed to rank as an alloying element whose presence is desirable in order to obtain certain advantages. In the presence of some alloy elements, especially chromium and aluminium, a slightly higher carbon content may be permitted in conjunction with high phosphorus, without detrimental effect; but in view of the tendency of phosphorus to segregate, it is important that the upper limit of phosphorus content usually observed in this country should not be increased except in steels of very low carbon content.

Martensite.*

THE nature of martensite is a problem which has occupied the attention of investigators ever since the recognition of this constituent in the micro-structure of steel, but no universally accepted view as to its nature may yet be said to exist. In summarising the position, Epstein[1] in his recent book on "The Alloys of Iron and Carbon" concludes that "martensite is a body-centred solid solution; as quenched it is tetragonal rather than cubic; the carbon or carbide is held in supersaturated solid solution, as indicated by the distention of the crystal lattice according to the carbon content." The hardness of martensite is ascribed by the same author "in the main to the dispersion or precipitation of carbon from the supersaturated solution by analogy with the hardening of the many other age-hardening alloys." He adds that the other factors which have been suggested as causes for the hardness of martensite, such as the allotropic change of gamma to alpha, the tetragonal structure, its needle-like or acicular structure, its fine grain size, the distortion of its lattice and the presence of internal strain, should be considered to be of only subsidiary importance.

This view, which expresses a widely recognised opinion, is challenged by the conclusions of A. von Vegesack,[2] derived from an extensive investigation based on the examination of six steels containing 0·70, 1·04, 1·13, and about 1·24 per cent. of carbon, with 0·16 to 0·38 per cent. silicon, and 0·16 to 0·54 per cent. manganese, which were quenched from 800 or 1100 deg. Cent. in brine and subsequently tempered at temperatures between 50 and 700 deg. Cent. in successive steps of 25 deg. by heating for 5 hours at each temperature. The specimens were ground, polished, and etched with 2 per cent. nitric acid in alcohol after each tempering treatment, and von Vegesack's opinions were based mainly on the results of microscopic examination, the important point about which was that, in spite of a general darkening of the specimen after low-temperature tempering, the actual martensite needles remained bright throughout. He therefore concluded that at

* A review and discussion of the ideas of A. von Vegesack.

no time during the hardening and tempering process does the martensite contain more carbon than the normal carbon content of α iron.

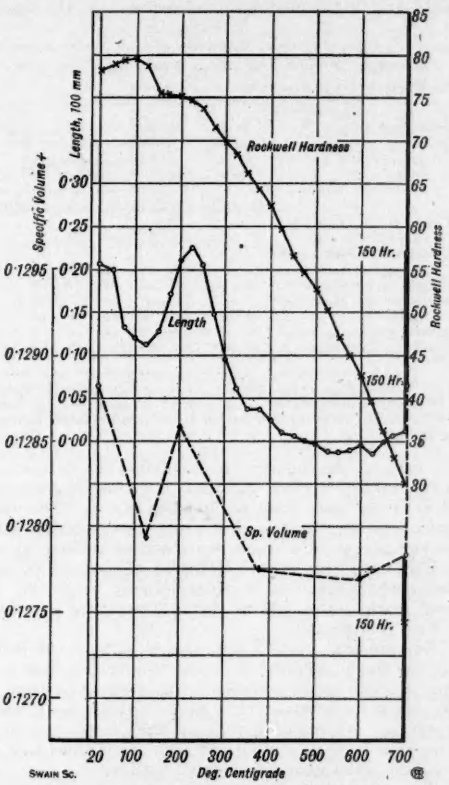
This opinion led to a lively discussion.[3] Hägg and Ohman,[4] who hold that as tetragonal martensite is tempered the carbon gradually precipitates from a supersaturated α iron solution, carried out a repeat experiment with a steel containing carbon 1·58, silicon 0·19, and manganese 0·32 per cent. quenched in water from 1100 deg. Cent., and tempered *in vacuo* at 400 deg. Cent. On etching with alcoholic nitric acid they found distinct darkening of the martensite needles, though after tempering at higher temperatures (*e.g.*, 675 deg. Cent.), the segregation of carbide into the boundaries left the needles again bright. They suggested that owing to oxidation the surfaces of von Vegesack's specimens had become preferentially decarburised at the martensite needles. This was refuted by von Vegesack, who pointed out that surface material was always removed by grinding, before microscopic examination, after each tempering. Nevertheless, it was not proved by repeated examination of a selected field that all the martensite needles remained bright. In fact, it appeared that only isolated martensite needles among those originally present may have remained bright, while the greater part of the section had darkened. Moreover, Hultgren,[5] on repeating a micro-examination, found that finely divided cementite could exist in the martensite needles, as already observed by Hägg and Ohman.

According to von Vegesack, martensite at the instant of formation contains only the normal carbon content of α iron. Where, then, does the excess carbon go ? The suggestion is that it is either precipitated in a finely dispersed form or remains dissolved in the residual austenite in the boundaries of the needles. Now, as is well known, there is much divergence of view as to how the carbon occurs in a hardened steel. For example, Wever and Naeser[6] concluded that the carbon, separating from martensite below 200 deg. Cent., exists in a dispersed state in α iron, and that on tempering above 200 to 240 deg. Cent. cementite forms. Von Vegesack, however, holds that the maximum amount of finely dispersed carbon is present in the quenched condition, and that the effect of tempering at low temperatures is to make the greater part of it redissolve in the retained austenite, which on tempering at a higher temperature is ultimately resolved into cementite and ferrite. On this view, the carbon content of the retained austenite must be much above the generally accepted saturation limit. Thus, assuming that a 0·95 per cent. carbon steel quenched from 1200 deg. Cent. in water contains about 14·5 per cent. of retained austenite (Maurer and Schroeter[7]), complete solution of the carbon would imply a solubility of carbon in austenite of 6·5 per cent. Ability to dissolve carbon in excess of the normal amount would, it is said, be conferred on the austenite by the high pressure to which it is subjected in the quenching operation, and by the precipitation of the carbon.

According to von Vegesack, only the carbon which has at some stage been dissolved in the residual austenite can form cementite. Any additional carbon present on quenching, which does not subsequently dissolve, cannot form cementite. In this connection, von Vegesack carried out by chemical and microscopical methods a number of experiments on the occurrence of temper carbon in his steels. He found that a steel containing 1·23 per cent. carbon after being quenched from 1100 deg. Cent. and tempered in stages up to 700 deg. Cent. contained 1·0 per cent. of temper carbon after the final tempering, while after 150 hours at 700 deg. the whole of the carbon was present as temper carbon. A steel with 1·04 per cent. carbon after similar treatments contained 0·03 and 0·35 per cent. respectively of temper carbon, while a 0·70 per cent. carbon steel contained only a trace. Less temper carbon was found not only with lower carbon content in the steels, but also with a reduced quenching temperature. This he explains on the grounds that the structure on hardening from a low quenching temperature, being finer, offers more opportunity of intimate contact between the carbon particles and the residual austenite, and thus increases the speed of re-solution of the carbon. Prolonged heating at 700 deg. Cent. produced no temper carbon in steels which had not been hardened. This part of von Vegesack's exposition seems to offer considerable difficulty. Whereas cementite in hypereutectoid steel is well known to break down when heated at 700 deg. Cent.,[8] it is hard to understand why temper carbon, if it represents elementary carbon surviving from the hardening process, will increase in amount on prolonging the period of heating at this temperature.

The experiments of von Vegesack were not confined to chemical and microscopical examination. He also made accurate determinations of length, specific volume, and Rockwell hardness of the steels he used, and found that tempering at 75–125 deg. caused a slight increase in hardness which then decreased with a halt in the fall at 200–250 deg. Length and specific volume decreased to a minimum at about 125 deg., increased to a maximum at about 250 deg., then fell again, first quickly to about 375 deg., then more slowly until 600 deg. was reached, after which there was a rise which became considerable on long tempering at 700 deg. Cent. The larger the increase, the more rapid was the decrease in hardness. The changes in the 1·04 per cent. carbon steel quenched from 1100 deg. are shown in the diagram. The maxima and minima were more pronounced in 1·24 per cent. carbon steel, less pronounced when the hardening temperature was lower and almost smoothed out when the carbon content of the steel was only 0·7 per cent. Von Vegesack explained these changes on the following lines. The first decrease in specific volume is due to absorption of carbon by the residual austenite. This relieves internal pressure, so that the austenite decomposes more readily into martensite and carbon, hence giving rise to the increase in specific volume at about 250 deg. Thereafter, on raising the tempering temperature, a state of equilibrium depending on temperature and internal pressure exists between the relatively carbon-free martensite, supersaturated austenite and carbon ; and subsequently there occurs the decomposition of retained austenite into α iron and cementite. The increase in specific volume above 600 deg. is attributed to the formation of temper carbon. Von Vegesack's interpretation of the sequence of physical changes does not agree with Enlund's results[9] on electrical resistivity, which shows an almost constant value on tempering up to 100 deg. Cent., gradual diminution from 100 to 240 deg. (carbon precipitation from martensite), a sharp drop at 250 to 300 deg. (change of austenite to martensite), followed by a slow reduction (coalescence of carbide). He recognises this, but explains[10] the fall in resistivity at 100 to 240 deg. (when carbon is supposed by him to be going into solution) by the accompanying release of internal stress.

Finally, some reference may be made to von Vegesack's reconciliation of his views with the evidence of X-ray analysis. It will be recalled that Hägg[11] found that on tempering a quenched 1·11 per cent. carbon steel at 80 deg. Cent. for up to 5760 min.[12] there was a gradual change in the axial ratio of the tetragonal lattice of martensite, but no final change to the cubic lattice, a steady state being realised when the carbon remaining in solution was apparently about 0·6 per cent. Carbon-containing martensite was tetragonal; cubic martensite, in Hägg's opinion, would contain very little carbon in solution. Von Vegesack asserts that all martensite is almost



Changes in Hardness, Length, and Specific Volume of Steel containing C 1·04, Si 0·18, Mn 0·17 per cent., Quenched in Brine from 1100 deg. Cent., and Tempered for 5 hours successively at each temperature.

carbon-free and makes the assumption that the tetragonal lattice is the result of marked deformation of the usual α iron lattice through the mechanical pressure exerted externally to the lattice as a result of carbon precipitation in the hardening process. On tempering, the axial ratios would gradually become so nearly 1 that the distinction between tetragonal and cubic martensite could no longer be recognised on account of the alleged uncertainty of the X-ray measurements, an uncertainty which was denied by Hägg and Ohman.

It is clear that von Vegesack's conception of the nature of martensite and of the hardening process introduces as many difficulties as it sets out to solve, and although this may be a characteristic of many individual discussions which have centred round this question in the past, few of them have offered such a complete break with the general line of thought on the subject. It is felt that the somewhat bold speculations which are made by von Vegesack will require a great deal more support from experimental evidence before they can shake the more generally accepted views summarised in the words of Epstein at the beginning of this article.

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[1] "The Alloys of Iron and Carbon," Vol. I, page 209.
[2] *Jernkontorets Annalen*, 1936, 120, 49–98; *Zeit. anorg. Chem.*, 1936, 227, 145–178.
[3] See *Stahl und Eisen*, 1936, 56, 1485–86.
[4] *Jernkontorets Annalen*, 1936, 120, 143–154, 294–299.
[5] *Jernkontorets Annalen*, 1936, 120, 304–5.
[6] *Mitt. K.W. Inst. Eisenforschung*, 1933, 5, 37–47.
[7] *Stahl und Eisen*, 1929, 49, 929–940.
[8] "The Alloys of Iron and Carbon," Vol. I, page 144.
[9] *Journ, Iron and Steel Inst.*, 1925 (1), 111, 305–314; *Jernkontorets Annalen*, 1936, 120, 226–232.
[10] *Jernkontorets Annalen*, 1936, 120, 232.
[11] *Journ., Iron and Steel Inst.*, 1934 (2), 130, 439–451; *Jernkontorets Annalen*, 1934, 118, 173–203.
[12] Not *hours*, as stated by Epstein, *op. cit.* page 214.

Chromium-Manganese Steels.

THE austenitic structure resulting from the presence of nickel in high chromium steel may also be induced by the addition of manganese. It is natural, therefore, that research should have been directed to the properties of high chromium-manganese steels, and that efforts should have been made to employ them in substitution for the more usual chromium-nickel austenitic steels.

The constitution of the iron corner of the iron-manganese-chromium diagram, up to 40 per cent. of manganese and 30 per cent. of chromium, has been explored by W. Köster.[1] With less than 12 per cent. of chromium, the structure and behaviour of the alloys was similar to that of the binary alloys of iron and manganese. With 12 to 28 per cent. of chromium, the alloys were found to lie within the α+γ region which extended parallel to the iron-manganese side. The properties of steels of this type were determined, and compared with those of chromium-nickel steels, by I. S. Günzburg, N. A. Alexandrova and L. S. Geldermann.[2]

The effect of nickel in promoting the formation of austenite was found to be twice as great as that of manganese. Austenite is produced by 8 per cent. of nickel, but requires at least 16 per cent. of manganese. Steels containing nickel 6, manganese 2 or manganese 7, copper 2 per cent., were also found to be austenitic. The chromium-manganese steels were similar in strength to chromium-nickel steels, but were somewhat inferior in corrosion resistance except against attack by sulphur-containing gases.

Investigations of the properties of chromium-manganese steel have recently been reported in papers by F. Brühl[3] and by M. Schmidt and H. Legat.[4]

[1] *Archiv für das Eisenhüttenwesen*, 1933-34, 7, 687–688.
[2] *Archiv für das Eisenhüttenwesen*, 1934–35, 8, 121–123.
[3] "Structure and Properties of Chromium-Manganese Steels with up to 1 per cent. C, 15 per cent Mn, and 30 per cent. Cr," *Archiv für das Eisenhüttenwesen*, 1936–37 (Dec., 1936), 10, 243–255.
[4] "Heat-Resisting Chromium-Manganese Steels," *Archiv für das Eisenhüttenwesen* 1936–37 (Jan., 1937), 10, 297–303.

Brühl worked on steels containing approximately 0·03, 0·25, 0·50, and 1·0 per cent. carbon, with 5, 10, and 15 per cent. of manganese and 15, 20, and 30 per cent. of chromium (silicon 0·3 to 0·8 per cent.), made in 50 kilos. melts in the high-frequency furnace.

They were forged at 1050–1080 deg. Cent. to 25 mm. round for mechanical tests, or 27 mm. by 7 mm. rectangular bar to provide material for micro-examination, hardness, and magnetic tests. The steels with 1 per cent. of carbon were the most difficult to forge, and the forging temperature of these steels had to be reduced by 50 deg. Steels with manganese 15 and chromium 30 per cent. were not forgeable whatever the carbon content; in fact, they disintegrated on being reheated for forging. The difficulty was overcome, however, by withdrawing the ingot from the mould before its temperature fell below 950 deg. Cent., soaking in a furnace at that temperature, and gradually heating to 1020 deg. The steel was forged at 1020 deg. Cent., the temperature never being allowed to drop below 900 deg., and the forging cooled in water.

Heat treatments for hardness and magnetic tests included quenching from 1000 deg., 1100 deg., or 1200 deg. Cent., followed by tempering for eight hours at 500 deg., 600 deg., 700 deg., or 800 deg. Cent. Corrosion, tensile, and notched bar impact test pieces were prepared in a similar way. Thermal arrest and dilatation curves were made on some compositions, and X-ray analysis carried out at the Kaiser-Wilhelm Institute under the direction of Drs. Wever and Möller.

Tofaute, Küttner, and Büttinghaus[5] found the limits of stability of the γ-region in iron-chromium-carbon alloys and indicated that the higher the chromium, the higher the minimum carbon content required to bring the alloy within the γ-region. Thus, at 1200 deg. Cent. and 20 per cent. of chromium the γ-region extended down to 0·6 per cent. carbon, with 15 per cent. to 0·3 per cent. carbon, but with 30 per cent. of chromium over 1·0 per cent. of carbon was required to bring the alloys into the α+γ region. By means of quenching experiments, Brühl found that the γ-region of the iron-chromium-carbon alloys was extended to a lower carbon content by the addition of manganese. To retain austenite in a 15 per cent. chromium steel, 0·3 per cent. of carbon is required, irrespective of the amount of manganese between 5 and 15 per cent. Between these limits the manganese contents were practically without influence on the limiting compositions of the constituents, which were determined by the chromium and carbon contents.

These conclusions are in accordance with the observations of Köster referred to above, which indicate that chromium content exerts the preponderant influence on the structure of the iron-manganese-chromium alloys. There is thus an essential difference between the action of manganese and that of nickel. Manganese has no direct austenite-forming action, but its influence is solely to reduce the critical cooling velocity and to render the austenite, once formed, more stable at low temperatures and under prolonged temperating treatments.

The steels, after quenching from 1200 deg. Cent., consisted of γ, γ+α, or α-solid solution, according to their carbon and chromium contents. The γ-solution hardened on reheating at 500–700 deg. Cent. owing to precipitation of carbide, and this was accompanied by some embrittlement. The main interest of the tempering experiments, however, was their indication of the behaviour of the α-solution, which was found to decompose into FeCr and austenite. The tempering of alloys with 30 per cent. of chromium gave rise to the formation of FeCr. This was shown both metallographically and by X-ray analysis, and accounts for the extreme brittleness of the alloys after tempering or slow cooling.

On some of the alloys which might be of practical importance full tensile and notched bar impact tests were made. The results of these are given in diagrams from which some typical data are extracted in Table I. In the first alloy mentioned in the table the carbon was insufficient to cause the formation of the γ-solution, and there was a marked separation of FeCr on tempering at 700 deg. Cent. The test piece was very brittle and broke prematurely with no elongation. In steels with higher carbon, there is hardening due to precipitation of carbide in the austenite. The fall in ductility on tempering is very marked throughout, but not so great as when FeCr is precipitated.

TABLE I.—*Effect of Carbon Content and of Heat Treatment on the Mechanical Properties of Alloys Containing* 15 *per cent. of Manganese and* 20 *per cent. of Chromium.*

Carbon, per cent.	0·15		0·38		0·59		0·82	
Silicon	0·54		0·78		0·86		0·79	
Manganese	14·6		15·4		14·4		15·1	
Chromium	19·9		20·5		19·4		20·6	
	Q.	Q.T.	Q.	Q.T.	Q.	Q.T.	Q.	Q.T.
*Yield point, tons per square inch	30	—	30	41	30	60	32	50
*Ultimate stress, tons per square inch ...	43	—	48	48	52	68	60	76
Elongation, per cent. ...	30	—	41	5	58	7	35	5
Reduction of area, per cent....	60	—	50	6	49	9	30	7
Impact figure, mkg./cm.²	21	0·4	10	2	15	2	7	2
Brinell hardness number	210	395	220	282	240	342	270	360

Q.: Heated 2 hours at 1200 deg. Cent. and quenched in water.
Q.T.: Quenched as above, reheated 16 hours at 700 deg. Cent.
* Converted from kg. per square mm. to the nearest ton per square inch.

Experiments showed that the weldability of steels having the possibility of useful application was not bad, but corrosion resistance was inferior to that of the usual chromium-nickel rustless steels, and much worse in the tempered than in the quenched steels. The resistance to scaling was also lower than that of the heat-resisting steels usually employed.

The heat-resisting properties were determined in greater detail by Schmidt and Legat,[4] who examined a greater range of low carbon alloys containing carbon under 0·2 per cent. with manganese up to 35 and chromium up to 30 per cent., and concluded that chromium-manganese steels could be used as substitutes for heat-resisting nickel-chromium steels within a limited temperature range. On the basis of previous knowledge and of a special study of the microstructure of selected alloys, they constructed a diagram (Fig. 1) representing the constitution of the alloys at 1100 deg. Cent. and at room temperature. At 1100 deg. Cent. the diagram is divided into three fields by the lines A B C and D E, viz., γ below A B C, α above D E, and α+γ between. On cooling to room temperature the ferrite in alloys above D E persists unchanged, as also does the austenite in alloys in the region C B B,′ while that in alloys to the left of the line B B′ breaks down partially giving

[5] *Archiv für das Eisenhüttenwesen*, 1935–36, 9, 607; see THE METALLURGIST, Aug., 1936, page 152.

martensite and austenite, or more completely in alloys of lower manganese content giving ferrite and martensite. The field above A C is traversed by the boundary of the compositions of alloys in which FeCr is produced by slow cooling or by annealing between 500 deg. and 850 deg. Cent. The line F *f* G represents this boundary as determined by forty hours' annealing at 570–660 deg. Cent. As a result of more prolonged annealing the line F *f* G is shifted to the left until it merges with B C. It is necessary

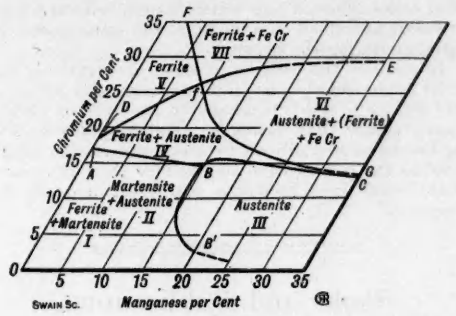


FIG. 1—*Constitution of Iron-Chromium-Manganese Alloys with 0·1 per cent. C. Sections at 20 deg. and 1100 deg. Cent.*

to avoid compositions above the line F *f* G on account of the embrittling effect of the separation of FeCr.

The amount of scaling which occurs in some typical steels (Table II) heated for seventy-two hours in a current of air is shown in Fig. 2. It was deduced that with a chromium content of 19 per cent. the

TABLE II.—*Percentage Composition of Steels used in Scaling Tests (Fig. 2).*

Steel.	C.	Si.	Mn.	Cr.	Ni.
A	0·21	0·95	8·0	19·0	1·0
B	0·16	0·58	15·7	16·8	1·3
D	0·15	1·59	1·2	25·1	17·2
E	0·18	0·02	16·3	14·2	—

steels have a sufficiently high resistance to scaling at 900 deg. Cent. They, however, show marked embrittlement on heating at 650 deg. Cent., and this increases on long exposure at that temperature. Thus, the impact figure of steel A (Table II), heated for 150 hours at 650 deg. Cent., fell to 1·7 mkg. per square centimetre. Therefore, if the steel is intended for use at 500–700 deg. Cent. it is desirable to select one in the austenitic range, *i.e.*, the chromium must be reduced to below 16 per cent. with 17 per cent. of manganese. The limit of practical scaling resistance of such a steel lies at 820–850 deg. Cent.

The silicon content of chromium-manganese steels should not be too high (and the same applies to aluminium) as it acts similarly to a higher chromium content only in a more pronounced manner. Schmidt and Legat's steel A contained 1 per cent. of nickel with advantage to its properties. The addition of 1 per cent. of copper to chromium-manganese steel is also said to lead to marked improvement. Good results are reported by A. B. Kinzel[6] to have been obtained with a steel containing carbon 0·1, chromium 18, manganese 8, and copper 1 per cent. Quenched from 1050 deg. Cent. it is composed of 60 per cent. austenite and 40 per cent. chromiferous ferrite. It is more easy to roll and to draw than a purely austenitic steel and has good physical properties, including resistance to corrosion. It is, however, subject to some intergranular corrosion, which cannot be avoided by the addition of a carbide-forming element. The addition of nickel, however, effects a further improvement, though this does not for all purposes persist up to complete replacement of the manganese. The presence of 4 per cent. manganese in a 14 : 10 chromium nickel steel was

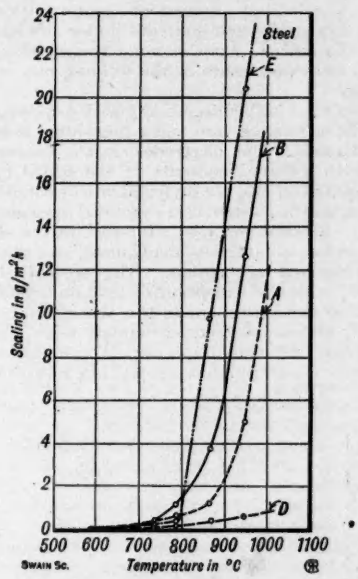


FIG. 2—*Scaling of Steels after 72 hours in Air at Different Temperatures (for composition, see Table II).*

found by Colbeck to lead to preservation of ductility and impact figure down to very low temperatures.

On the whole, it seems that the chromium-manganese steels in which all the nickel of a chromium-nickel steel is replaced by manganese, will have only a very limited field of application on account of their relatively poor properties compared with those of the chromium-nickel steels.

[6] *Metals and Alloys*, December, 1936, 7, M A 604.

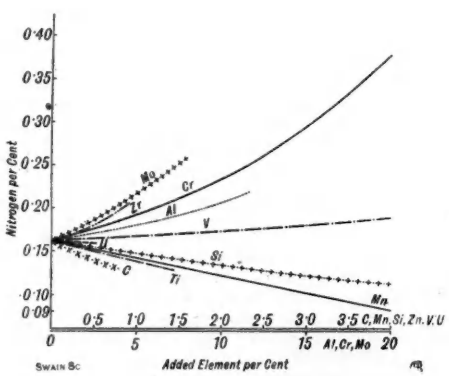
Nitrogen Absorption in Arc Welding.

IT has long been recognised that to secure the best mechanical properties in welds the deposited weld metal should contain the lowest possible quantities of oxygen and nitrogen. The nitrogen content of welds varies considerably according to the nature of the welding rod and degree of protection afforded by the flux. Studies by Portevin and Séférian on

nitrogen in relation to welding are well known. A version in French of a paper given by them at the Iron and Steel Institute Welding Symposium, 1935, has recently appeared,* but to it another section has been added† dealing with the influence of special elements in steel on the absorption of nitrogen. This part of the work is also described in a paper by the same authors,‡ entitled "Influence of Addition Elements in Steel on the Absorption of Nitrogen during Fusion in the Arc."

With a given electrode coating there is usually a corresponding nitrogen concentration which is quickly attained and thereafter remains constant. For example, with the so-called neutral flux, formed from a mixture of felspar and oxide of iron (Fe_2O_3 47, SiO_2 40, and Al_2O_3 8 per cent.), this limit is 0·04 per cent., a relatively dangerous concentration. To obtain lower nitrogen contents under similar conditions, the authors have in mind the possible use of special elements, either in the welding rod or in the coating.

Alloys of as high a degree of purity as possible were made from Swedish iron and a ferro-alloy rich in the special metal. The electrodes, in the form of bare wire with different contents of the added element, were melted in the arc under identical conditions of current, and the weld metal examined microscopically and by chemical analyses. Increase in the elements common in steel (carbon, manganese, and silicon) led to diminished absorption. The special elements studied, with the exception of titanium, tended, on the other hand, to increase the absorption. Molybdenum, zirconium, and chromium have the most pronounced influence in this respect, as indicated in the diagram. Under the microscope, TiN had the colour



of yellow brass, ZrN was straw coloured, VN pale grey with a violet glint, U_3N_2 pinkish yellow. All were polyhedral, except uranium nitride. Aluminium nitride may possibly have been recognised, but no new constituents appeared in the alloys containing manganese, silicon, chromium, or molybdenum. Steels containing different amounts of silicon, carbon, manganese, and chromium were also studied in the form of electrodes coated with flux of the neutral type indicated above, and the simultaneous effect of the addition element and of relative dimensions of rod and coating was determined. For any given relation of coating to rod the tendency was for the nitrogen absorption to be reduced by increase of manganese, and increased by increase of chromium content.

No reference is made to the work of Schuster,§ though as regards carbon, silicon, and manganese the work serves to confirm his conclusions. Portevin and Séférian's work leads them to suppose that nitrogen absorption may be an advantage in the ferritic chromium steels, since what would otherwise be a coarse-grained and brittle structure in the zone of fusion is refined as a result of the introduction of nitrogen during the fusion.

In spite of the interest of these observations, the most important chemical influence adversely affecting the quality of welds is oxygen in the form of ferrous oxide, rather than nitrogen; and in practice it is by the use of specially coated electrodes and by attention to the constitution of the flux that difficulties with both these elements can most suitably be overcome.

* *Revue de Métallurgie*, Dec., 1936, 33, 705.
† *Ibid.*, Feb., 1937.
‡ *Comptes Rendus*, 1937, 204, 48.

Books and Publications.

Hardness of Metals. By F. C. LEA. Medium 8vo, pp. vi+141. London: Charles Griffin and Co., Ltd. 12s. 6d.

IN the course of his extensive researches on the properties of materials, Professor Lea has had frequent occasion to use that most convenient of all tests for checking uniformity and following the changes which occur during heat treatment or mechanical deformation—the hardness test. He has used most forms of the test and so is in a position to indicate, from his own experience, the relationship between the results obtained by different methods, the extent of the errors involved in each of them, the degree of accuracy with which hardness can be measured, and the definiteness with which it should be specified.

The book contains a description of the methods used, which include the Vickers diamond pyramid hardness test, the Firth hardometer, various types of Brinell ball hardness machines, the cone indentor machines (Rockwell and Avery), the Herbert pendulum hardness test and the Scleroscope.

The theory of each method is briefly discussed and the effect of variations from the standard conditions of the test are illustrated by actual experimental results. The final comparison of different types of test is expressed in the form of curves and charts from which the equivalent hardness numbers on the different scales of hardness can be read off. These results are valuable as illustrating the careful work of one laboratory, but the drawback of such centralisation is that it does not afford a check on the idiosyncrasies of individual machines, and that it limits the range of materials tested. Several more such series, carried out in other establishments, must precede the final acceptance of either a conversion curve or a figure expressing probable error.

The hardness test is so widely used in metallurgical and engineering practice that there can hardly be any laboratory dealing with strength of materials which does not possess some means of conversion of one hardness scale into another, and one of the

§ *Journ.*, Iron and Steel Inst., 1932 (1), 125, 143.

first things that the reader of this book will do is to check his own conversion curves against those given by Professor Lea. For the purpose of illustration it may be interesting to go rather far afield for a comparison, and so there are recorded in the accompanying table the figures read from charts or curves given by Professor Lea and by the American Society for Metals, in the *Metals Handbook*, 1936. A comparison may also be made with the curves reproduced in THE METALLURGIST (December, 1936, pages 188–190). There is a good deal of evidence to show that over the D.P.H. range 200 to 350 the Brinell hardness is lower than the American tables show, but perhaps not quite so low as Professor Lea's figures indicate.

"*Hardness of Metals.*" *Comparison of Professor Lea's Equivalent Values and those of the American Society for Metals.*

Vickers 136 deg. diamond pyramid hardness, 50 kilos. load, 12 sec.	Brinell hardness, 10 mm. ball, 3000 kilos. load, 30 sec.		Rockwell C.	Rockwell C.	Avery.	Shore Scleroscope.	
			120 deg. cone, 150 kilos. load, 10 sec.				
	A.S.M.	Lea.	A.S.M.	Lea.	Lea.	A.S.M.	Lea.
200	200	190	14	12	11	29	22
300	297	280	31	34	29	42	36
400	384	370	41	44	40	54	47
500	466	460	47	52	49	63	55½
600	534	544	53	57½	55	72	62
700	593	595	58	62	58½	80	68½
800	—	—	61	65½	62½	86	74½

The Rockwell and Avery cone indentation tests, in both of which a 120 deg. cone and a 150 kilo. load for ten seconds are employed, might be expected to give identical results. The Avery machine, however, gave consistently lower results than the Rockwell. This was not due to the use of a worn diamond, though it was shown that the hardness number obtained with a worn diamond was considerably lower than with a new diamond. The American Rockwell curve agrees most nearly with Professor Lea's results on the Avery machine. The most serious discrepancy is between the Shore Scleroscope results. It is noted that Professor Lea's values are based on results obtained for the most part on specimens not more than ¼in. thick, and it may be that the low mass of the specimen tended to reduce the Scleroscope reading. Another point is that the Scleroscope equivalent of a given D.P.H. number is very dependent on the nature of the material, one with a high elastic ratio giving a greater rebound for the same diamond pyramid hardness than one with a low elastic ratio. Unfortunately, the description of some of the specimens is very brief, *e.g.*, "vanadium steel" and compositions and full mechanical tests are frequently not given.

Professor Lea's book also contains a discussion of hardness tests at high temperatures carried out (1) in the ordinary way, the specimen being initially heated, and (2) by impact. In both cases the curves follow the general shape of the ultimate stress curves in short-time tensile tests, although in the dynamic hardness tests the temperature of maximum value is displaced from 250 deg. to 200 deg. Cent. Reference is made to the softening of the steel ball and its effect. At present the use of cemented carbide balls would enable static tests (which have the advantage of not involving any conversion factor to obtain the Brinell hardness number) to be made at temperatures up to 800 deg. Cent.

Strain hardening and the distribution of strain hardening in tensile and other test pieces, and in hot and cold-drawn wire, are also discussed. There is a brief reference to embrittlement produced by prolonged heating of certain alloy steels at raised temperatures and to quench-age-embrittlement. In this part of the book the "Tensile Strength: Hardness" ratio is frequently given, but the number of instances for materials other than steel are too few to make the figures for the non-ferrous metals and alloys very significant. For steel, except when there was obvious premature failure in the tensile test, or lack of uniformity (as in free-cutting steel) the value of the ratio varied from 0·20 to 0·23, in heat-treated 0·4 per cent. carbon steel from 0·204 to 0·22.

Professor Lea has written an interesting monograph on the measurement of the hardness of metals, but one cannot help feeling that the mass of detailed results contained in many of the fifty-one tables which form so large a part of the book, and sometimes extend even to duplicate ocular readings of the individual hardness impressions, are not essential. In other words, no one is in such a good position as Professor Lea to assess critically the data presented, and a much shorter summary of the work would have been not only adequate, but even more helpful to the reader, in presenting concisely the judgment of an eminent investigator, of whose work all interested in hardness testing will be anxious to make full use.

Fertigungstechnik und Güte abbrenngeschweisster Verbindungen. By Dr.-Ing. HANS KILGER. Brunswick: Frederick Vieweg and Son. 1936. Pp. 128, 75 illustrations, 15 tables. Price R.M. 7·50.

DR. KILGER has added another, and very readable, book to the already large number of publications upon resistance welding. Devoted exclusively to the theory and practice of flash welding, it is based primarily upon research carried out at the works of Siemens-Schuckert between 1931 and 1934. Among the virtues of the process, recorded in the preface, is the possibility it possesses of being turned to good account in the campaign for economic self-sufficiency; the author is careful to point out that in this connection full use is not made of the wide applicability and efficiency of the method.

The first thirty-eight pages are concerned with the theoretical and practical aspects of the process itself, and open with a somewhat brief description of the plant employed, the basic principles embodied in such plant, and the general technique of flash welding. A few more illustrations showing various types of

machine, and possibly some diagrams of the significant units composing the machine would have enhanced the value of this section of the book. The photographs on page 20, depicting the effect upon the surfaces to be welded of various quantities of energy, are very apt, and the tendency to porosity caused by unduly heavy energy input is very well shown. In addition, an excellent series of photomicrographs illustrates the structural alterations, resultant upon welding, which are dealt with in the succeeding section. In Table III, however, giving the analyses of the steels used by the author in his experiments, it will be noticed that in no case does the sulphur content exceed about 0·068 per cent. The omission of high-sulphur steels will be referred to again.

From pages 49 to 89 an excellent account is given of the results of tensile, hardness, and fatigue tests upon welded joints in various steels, but the section upon faulty welds which follows, while well illustrated, hardly conveys to the reader the importance of the time factor in the carrying out of the work, nor is sufficient emphasis laid upon the influence of section, sulphur content, and the capacity of the machine employed. These factors are of vital importance in practice, and to a very large extent must determine the type of machine or machines used; decidedly they will affect the choice of a hand-operated or fully automatic plant. As noted previously, no work was done on the higher sulphur steels; though with increasing concentrations of inclusions having a relatively low melting point the time of processing and the applied pressures become increasingly important, and when larger sections are being dealt with, conditions, in regard to these factors, become still more stringent.

The final pages cover the effect of the technique employed upon the quality of the joint and the effect of size and shape of section, an excellent bibliography and summary of patents forming the conclusion of the book.

Duralumin and its Heat Treatment. By P. LITHERLAND TEED. Medium 8vo. Pp. ix+116. London: Charles Griffin and Co., Ltd. 1937. 12s. 6d.

THIS book has been written by Major Teed, of Vickers (Aviation), Ltd., Weybridge, Surrey, not, he says in the Preface, for metallurgists, but for those engineers whose task may be facilitated by the use of duralumin, an alloy whose combination of relatively high mechanical properties with correspondingly low specific gravity renders it of great importance for an ever-increasing number of purposes. But the class to whom the book is dedicated, viz., "all who have suffered from the discreet silence or indiscreet inaccuracy of works of reference with regard to the heat treatment of duralumin," must contain a considerable number of metallurgists, who will be no less grateful than the engineers to the author for placing at their disposal the results of this very long series of carefully planned experiments, intended to exemplify the effects of a great variety of treatments on alloys, all of which fall within the fairly narrow limits of composition laid down in the British Standard Specifications for Duralumin for aeronautical purposes. The treatments investigated include "annealing," designed to put the alloy into a condition such as to allow the highest degree of plastic deformation to be carried out during the longest possible period following this treatment; "normalising," the solution treatment, intended to give temporary ductility for a short period and slow development of improved mechanical properties during subsequent ageing, the course of the age-hardening process itself and ageing under the influence of plastic deformation applied immediately after the "normalising" treatment. In all these the variables of time, temperature, and rate of cooling from that temperature were fully explored. The mechanical properties measured are clearly defined without ambiguity; for example, the term "yield point" is avoided, and the loads giving permanent extensions of 0·1, 0·2, and 0·3 per cent. are recorded. It would have been useful if the conditions of measurement of Young's modulus had been more explicitly stated, so that some estimate could be formed of its accuracy. The values for the cast material lie between 8·4 and 9·0, and those of the forged and heat-treated alloy mostly between 10 and 11 million lb. per square inch, but it is not possible to trace any effect of heat treatment on the results.

A great variety of products was investigated, from thin sheet to 5in. diameter bar, but throughout the book mechanical properties are the criterion by which the alloys are judged. Some reference to the effect of the various modifications of heat treatment on corrodibility or susceptibility to intercrystalline attack would have been welcome. The stove-enamelling process is, however, discussed, and a clear warning given that the process of stove enamelling, as at present carried out, cannot be applied without some loss of mechanical properties; consequently, its employment must have the inevitable effect of reducing the factor of safety of any stressed part to which it is applied.

Useful practical notes on the heat treatment of duralumin, the lay-out of the heat-treatment room, and on the design and use of salt baths are contained in the last chapter.

The book contains an adequate account of the development of duralumin and of the functions of the constituents of the alloy. There is also a short discussion of the theory of age hardening, which, however, is not complete enough to be entirely satisfactory. Nor perhaps could it be so in the present state of knowledge; at any rate, the author seems to hold that opinion, for after sowing a few "tares in the field of the emphatic theorist," he hastens on to give results which "are very definitely concerned with ordinary engineering practice."

Although many books and papers have been written on the subject of aluminium alloys and duralumin in particular, no such collection of strictly comparable tests on a single type of alloy is available elsewhere, and by the publication of this book Major Teed has made a substantial contribution to the advancement of the satisfactory production and wider use of duralumin.